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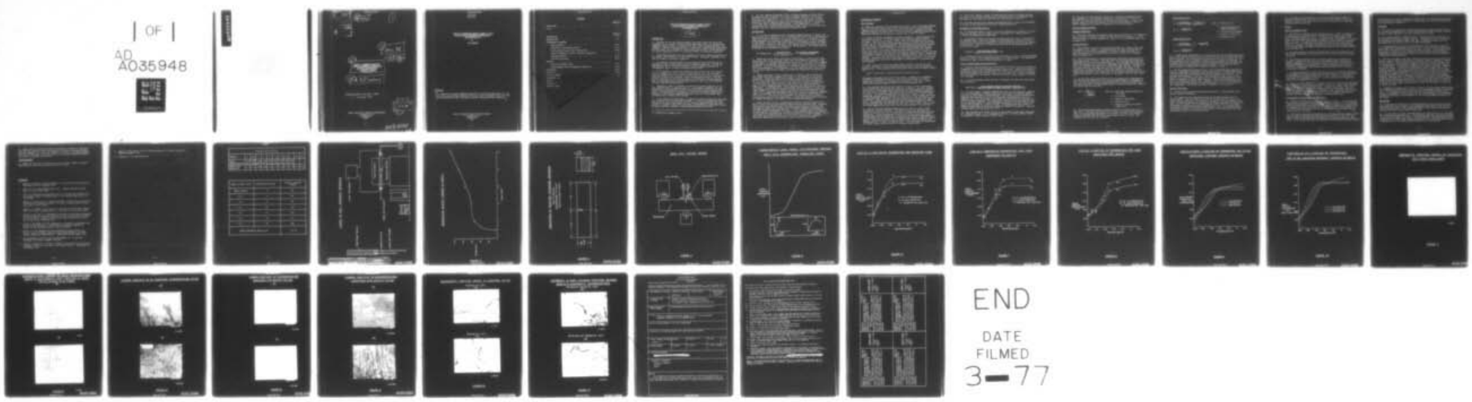

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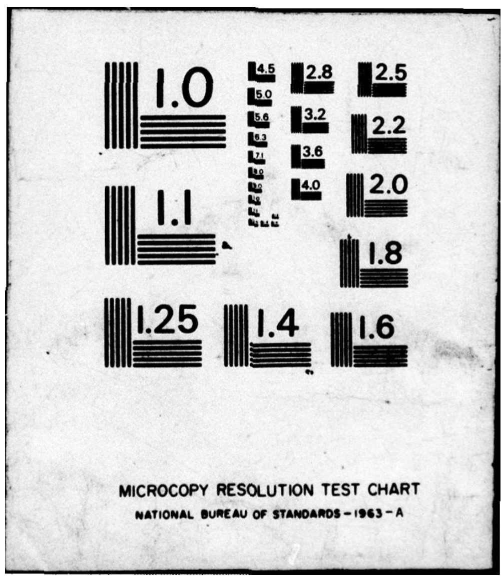
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EFFECT OF PERCENTAGE BAINITE FORMED AT 400 C ON THE FRACTURE TOUGHNESS PROPERTIES OF AS-QUENCHED HY130

(Item 6E5)

10 A. J. Caudrey

Naval Construction Research Establishment
St Leonard's Hill
Dunfermline
Fife

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ABSTRACT

The comparative fracture toughness properties of bainite and martensite in a low alloy high yield strength structural steel have been investigated using the crack opening displacement (COD) technique of general yielding fracture mechanics.

Naval Construction Research Establishment
St Leonard's Hill
Dunfermline
Fife

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INTRODUCTION

1. The need for low cost high performance structural steels has led to the development of low carbon, low alloy steels which when used in the quenched and tempered condition exhibit a combination of high tensile strength and high notch toughness. However, many applications of these steels require welded construction and the thermal cycles imposed by welding processes can alter the microstructure of a steel significantly and in so doing may modify its mechanical properties.

2. Welded structures must be viewed essentially as three component composites of weld metal, plate material, and heat affected zone. Inadequate toughness in any one is a threat to structural integrity.

3. For practical testing purposes the weld metal and plate material may be isolated in order to obtain their fracture toughness properties. However the heat affected zone of the parent material is not so readily separated for investigation due to its small non-uniform curved shape.

4. It is possible to introduce an artifical crack in the form of a machined notch or fatigue crack into a real weld heat affected zone, but during the fracture toughness test, the fracture initiating from the notch tip immediately moves out of the area of interest, namely the heat affected zone. Also the plastic deformation at the notch tip may not be confined to the heat affected zone, but may spread into the surrounding material, which will have different fracture toughness characteristics.

5. Therefore in order to determine the fracture toughness properties of the heat affected zone as a separate entity it is necessary to reproduce the heat affected zone microstructure uniformly throughout the fracture toughness testpiece. The microstructure occurring in a real weld heat affected zone may be simulated in the fracture toughness testpiece by use of a Weld Thermal Cycle Simulator (1)* or furnace heat treatment.

6. It was decided for this project to simulate the microstructure of a real weld heat affected zone by furnace heat treatment. To obtain the appropriate microstructures the blanks were heated into the austenite range and were then water quenched to produce a martensitic structure, or isothermally transformed followed by water quenching to produce a mixed martensite/bainite structure.

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*( ) References on pages 13 and 14

7. The crack opening displacement (COD) technique of general yielding fracture mechanics was used to evaluate the fracture toughness properties of the microstructures under examination. It is known that HY130 has good notch toughness characteristics especially in 10 mm thick sections, so a high degree of plastic deformation could be expected before failure in specimens tested at temperatures on the upper shelf of the transition curve. As the large amount of plastic deformation prior to failure would invalidate a linear elastic fracture mechanics approach, the crack opening displacement technique was used.

PREVIOUS WORK

8. In the case of hardenable steels containing moderate amounts of carbon, the rapid cooling rates associated with the ordinary weld thermal cycles generally cause formation of brittle martensite in the heat affected zone. The presence of high carbon martensite is invariably accompanied by a deterioration of notch toughness in this region. Since post weld heat treatment is not always possible, the alternative solution requires a reduction of the cooling rates after welding by either preheating or by increasing the energy input during welding. The energy input may be defined by the following relationship:

$$\text{Arc energy input} = \frac{\text{arc power (kJ)}}{\text{travel speed(mm s}^{-1})} = \frac{\text{arc voltage x arc current}}{\text{arc travel speed}}$$

9. However for alloy steels containing low percentages of carbon, notch toughness may be high in the as-welded state, even in the absence of preheat or high energy inputs. It has been shown by Aborn and Doty (2), (3), that notch toughness is retained in the quenched condition as a result of the formation of low carbon martensite. The low carbon martensite is conditioned while cooling by a process known as autotempering. Autotempering is the self tempering which occurs during the cooling of these steels through the temperature range below the relatively high M_s.

10. Nippes and Sibley (4) have verified that T-1 steel in the as-welded condition has a high notch toughness as a result of the formation of low carbon martensite. It has been postulated by Nippes, Savage and Allio (5) that preheat or high energy inputs are in fact detrimental to the notch toughness of T-1 steel since both factors lower cooling rates after welding and consequently tend to produce microstructures other than that of low carbon martensite. Nippes et al (5) prepared Charpy V notch specimens from simulated heat affected zone blanks. It was found that a 260°C preheat before welding at 47 kJ/in with a peak temperature of 1316°C, raised the ductility transition temperature from -95°C to -3°C. Increasing the energy input from 47 to 75.6 kJ/in had an analogous effect upon toughness by increasing the ductility transition temperature from -95°C to -27°C. This loss in notch toughness was attributed to the change in microstructure from low carbon martensite to a mixed martensite/bainite structure in a matrix of ferrite, as a result of slow cooling after welding.

11. Dolby and Knott (6) have investigated the effect of bainite formation as a result of slow cooling after welding in HY80 weld heat affected zones. Simulated heat affected zone microstructures containing varying amounts of bainite were developed by furnace heat treatment and the fracture toughness was measured by the crack opening displacement technique. It was concluded that the introduction of upper bainite into martensite of constant prior austenite grain size progressively lowered the toughness when fracture initiation was by quasi-cleavage. It was also found that bainite contents up to 60% did not influence the toughness of martensite significantly when initiation was by microvoid coalescence.

EXPERIMENTAL PROCEDURE

Heat Treatment

12. Thirty-one specimen blanks each measuring 10 mm x 10 mm x 60 mm were machined from $1\frac{1}{2}$" thick HY130, plate number 5922. The material was in the quenched and tempered condition having been austenitised at about 850°C followed by a water quench and tempering at about 550°C. The HY130 chemical composition specification MIL-S-24371 (Ships) and the analysis of plate 5922 is shown in Table 1.

13. Details of furnace arrangement are shown in Figure 1. Specimen blanks were individually heated in a 3 kW tube furnace containing argon gas to prevent oxidation. The furnace was held at 1350°C ± 5°C, and the specimens were heated to 1200°C ± 5°C in 140 secs giving a heating rate of 8.6°C/sec, from which temperature they were immediately quenched. The as-quenched specimens were quenched into cold water. The isothermally heat treated specimens were quenched into a lead bath at 400°C ± 5°C. The time of isothermal heat treatment was measured with a stop watch. The furnace temperature was measured by a Pt/Pt + 10% Rh thermocouple the output from which was monitored on a Rikadenki 6 Channel Recorder. The specimen temperature was measured by a sheathed Chromel v Alumel thermocouple inserted into a hole 5 mm deep and 4.5 mm in diameter drilled into the end of the specimen. The output from this thermocouple was again monitored on the Rikadenki 6 Channel Recorder, and from the chart speed of 50 mm/minute, the heating rates were obtained.

14. When cooling steel from high temperatures, austenite only starts transforming to martensite when the temperature falls below a critical temperature, denoted by M_s. The M_s temperature was calculated using Steven and Haynes formula (7):

$$M_s(^{\circ}C) = 561-474(\%\ C)-33(\%\ Mn)-17(\%\ Ni)-17(\%\ Cr)-21(\%\ Mo)$$

This gave a calculated value of 371°C for the M_s. Examination of isothermal transformation diagrams of plates at the top and bottom ends of the HY130 composition showed that an isothermal heat treatment at 400°C for 5 minutes should give high percentages of bainite in the microstructure (8).

15. Seven specimen blanks were heat treated to obtain a curve of percentage bainite formed at 400°C for various times up to 5 minutes. The first specimen was heated to a peak temperature of 1200°C and was then immediately quenched into cold water to produce a fully martensitic structure with 0% transformation to bainite. The second specimen blank was heated to a peak of 1200°C and was immediately quenched into a lead bath at 400°C for 30 seconds during which time it was continuously agitated to reach the temperature of the lead bath as soon as possible. After 30 seconds, the specimen was quickly quenched into cold water. The remaining five specimens were each individually heated to 1200°C, and were quenched into the lead bath at 400°C for 1 min, 2 mins, 3 mins, 4 mins and 5 mins respectively after which time they were quenched into cold water.

16. Each of the seven specimen blanks were cut in half on a water cooled diamond disc cutting off machine. One half of each of the specimens was mounted in Bakelite with the cut face ready for preparation for metallographic examination. Initial grinding was carried out on water lubricated silicon carbide papers from 120 grade down to 600 grade, the specimen being washed in flowing water, and the direction of polishing being rotated through 90° between papers. Diamond polishing followed on Dialap Fluid lubricated Simplex pads impregnated with 6 μm and 1 μm diamond paste. Final polishing was accomplished on a water soaked metron pad with 0.05 μm gamma alumina powder.

17. After final washing in water the specimens were etched in 2% Nital solution for between 5-10 seconds, depending upon the microstructure to be examined. The specimens were then washed in water and dried with methylated spirits.

18. The seven specimens were examined metallographically to obtain the percentage of bainite in each specimen and the prior austenite grain size.

Estimation of percentage bainite

19. The percentage bainite in each of the seven specimens was estimated by a point counting technique using an Olympus optical microscope with a magnification of x 320, and an eyepiece with cross wires.

20. The specimen surface was traversed at random using the x and y adjustments on the specimen table. After each movement the microstructure coincident with the centre of the cross wires was examined to decide whether the structure was bainitic or martensitic. Initially 250 fields were taken and the percentage bainite was given by the formula:

$$\%\ \text{bainite} = \frac{\text{No of bainite counts}}{\text{Total No of counts (250)}} \times 100$$

21. A further two sets of 250 fields each were carried out and the final percentage bainite was taken as the average percentage of the three figures. The results are tabulated in Table 1 and are shown graphically in Figure 2.

Prior austenite grain size determination

22. The prior austenite grain size was measured on a Vickers optical microscope at a magnification of x 450, using Xenon vapour illumination. Conventional grain size measurement etchants such as SASPA NANSA were found to be ineffective in etching the austenite grain size in this steel, but it was found that the grain boundaries were well defined in the specimen containing 96% bainite after etching in 2% Nital for 7 seconds.

23. The linear intercept method was used to determine the grain size where grain size is given by:

$$\text{Grain Size} = \frac{\text{Total magnified length of specimen examined}}{\text{Total No of grain boundary intercepts x magnification}}$$

24. The total length of specimen examined was found by multiplying the length of a 10 cm line engraved on a ground glass screen onto which the specimen image is projected, by the number of fields examined. The number of grain boundary intercepts per field was the sum of all the intercepts coinciding with the 10 cm engraved line. The total number of intercepts was the sum of the intercepts for all of the fields examined. The fields were selected at random across the whole of the specimen surface to avoid discrepancies caused by fine grain sizes at the edge of the specimen due to faster cooling. The specimen was also rotated at random between fields to reduce any orientation effects of the grains. The prior austenite grain size was found to be 50 µm.

25. From the graph of % bainite vs time it was decided to heat treat three sets of 8 specimens to give 1. an as-quenched structure, 2. bainite formed at 400°C/30 secs and 3. bainite formed at 400°C/5 mins.

26. The point counting technique showed the as-quenched microstructure to be fully martensitic. The specimens isothermally transformed at 400°C/30 secs and 400°C for 5 minutes formed 4% and 96% bainite respectively. A COD vs temperature transition curve was produced for each set of specimens, in order to obtain a measure of the relative toughnesses of each microstructure.

Fracture Toughness Testing

Specimen Preparation

27. After heat treatment the specimen blanks were machined down to the dimensions shown in Figure 3 in order to remove the lead coating the specimens. A 0.006" wide notch was cut into the specimens to a depth of 1/3 the specimen width, with a diamond disc.

Testing Procedure

28. Each specimen was tested in three point bending over a span of four times the specimen width, Figure 4. Testing procedure was based on the guidelines laid down in the Draft for Development of Fracture Toughness Testing (9). A Mand Servohydraulic testing machine was used for all tests with a crosshead rate of 1 mm/min (10).

29. The COD was measured by a double cantilever displacement gauge mounted on knife edges across the notch. The output from the clip gauge was plotted autographically on the x axis of a Bryans plotter. The clip gauge was calibrated with a jig (measuring to 0.0001" on a dial gauge) which allowed the gauge to be opened in increments corresponding to known knife edge openings. The value of the output from the clip gauge for each known increment was marked on the plotter and a calibration curve of knife edge opening as a function of clip gauge output was drawn. The output from a 50 kN load cell was monitored on the y axis of the Bryans plotter, thus giving an autographic record of load versus COD.

30. The COD values were calculated from the clip gauge opening displacements using both methods described in the British Standard Draft for Development 19 (9). The first equation derived from experimental calibrations assumes that deformation occurs by a hinge mechanism about a centre of rotation at a depth of 0.33 (w-a) below the crack tip:

$$\mathrm{COD} = \frac{V_g}{1 + \left[\dfrac{a+z}{r(w-a)}\right]}$$

where Vg = knife edge opening measured by clip gauge.

a = notch depth.

z = height of knife edges.

w = specimen width.

r = centre of rotation of plastic hinge (0.33).

31. The second method proposed in DD19 is based on the theoretical work of Wells (11). This treats the relationship between clip gauge opening displacement and crack opening displacement in two parts; the first, up to general yield, is parabolic and the second, beyond general yield is linear. The equations for each part are as follows:

Below general yield

$$\delta_c = \frac{0.45(w-a)}{0.45w + 0.55a+z} \left[\frac{V_c{}^2 E}{4\gamma\sigma y w(1-\nu^2)}\right] \quad \text{where } \gamma = V^1 E/\sigma y w(1-\nu^2).$$

V^1 = limiting elastic clip gauge displacement.

$$\text{for } V_c < \frac{2\gamma\sigma y w(1-\nu^2)}{E}$$

σy = material yield strength.

E = modulus of elasticity.

ν = Poisson's ratio.

Above general yield

$$\delta_c = \frac{-0.45(w-a)}{0.45w + 0.55a+z} \left[V_c - \frac{\gamma\sigma y w(1-\nu^2)}{E}\right]$$

$$\text{For } V_c \geq \frac{2\gamma\sigma y w(1-\nu^2)}{E}.$$

32. At temperatures corresponding to the lower shelf of the COD temperature transition curve, unstable fracture occurs in HY130 with very little plastic deformation prior to failure. The load versus COD record for such a test is normally of the form typified by curve 1 in Figure 5, where the COD at fracture is measured. On the upper shelf of the transition curve for HY130 appreciable plastic deformation occurs at the crack tip followed by ductile tearing. In this case the COD at first attainment of maximum load is measured, Figure 5, curve 2.

33. For test temperatures between 0°C and -70°C the specimens were immersed in a cold bath containing Iso-Pentane and solid carbon dioxide. Temperatures between -70°C and -196°C were attained by using liquid nitrogen in the cold bath. Temperatures were measured by a Chromel v Alumel thermocouple spot welded to the specimen in the vicinity of the notch, the temperatures being continuously monitored on a Rikadenki 6 Channel Chart Recorder. Control was estimated at ± 2°C down to -70°C and ± 5°C below -70°C. Transition curves for the three heat treatments were obtained by plotting COD (both Wells and r = 0.33), as a function of test temperature. The results are shown in Figures 6, 7, 8, 9 and 10.

Electron Microscopy

34. Polished and etched microsections were examined in a Philips EM300 transmission electron microscope.

35. Self shadowed carbon replicas were made by evaporating carbon onto the etched surface. This was carried out under vacuum (less than 10^{-5} Torr) in an Edwards Vacuum Coating Unit, by striking an arc between two spectrographic carbon rods. The carbon replicas were stripped by alternately immersing in a mixture of 10% bromine in methanol, and electro polishing at 20V, 0.6A in an electrolyte consisting of 10% perchloric acid in glacial acetic acid. The stripped replica was initially washed in a mixture of 20% methanol in distilled water, and final washing was carried out in distilled water.

36. The replicas were mounted on 3 mm diameter copper grids, dried, and examined in the electron microscope at an accelerating voltage of 80kV. The fracture surfaces of the three microstructures were examined in a Cambridge scanning electron microscope.

RESULTS

Fracture Toughness Tests

37. When HY130 is quenched from the austenite region to 400°C, transformation to bainite occurs; 4% bainite being obtained after 30 seconds; 96% bainite after 5 minutes, Figures 11 and 12. It appears likely from Figure 2 that there is an incubation time of about 25 seconds before transformation to bainite begins to occur. The prior austenite grains were clearly seen at a magnification of x 320 in the specimens containing 96% bainite, after etching in 2% Nital for 7 seconds. The grain size obtained by the linear intercept method of 50 μm falls within the range of grain sizes found in real weld heat affected zones in HY130 (12).

38. Increasing amounts of bainite progressively lowered the hardnesses from 399 VPN for the martensitic structure to 338 VPN for the structure containing 96% bainite.

39. The fracture toughness results are shown graphically in Figures 6-10 inclusive. Figures 6, 7 and 8 show the COD results of each test plotted against temperature for the three microstructures examined. All three microstructures exhibited a pronounced drop in crack opening displacement with reduction in temperature, indicating a fracture transition from a ductile to brittle mode of failure.

40. It can be seen that the correlation between the crack opening displacement as calculated by the Wells formula, and the experimental ($r' = 0.33$) formula is quite accurate at COD levels below about 0.07 mm. However the correlation diminishes with increasing COD values; the Wells formula value always being higher than the experimental value.

41. There is little difference in the upper shelf COD values for the martensitic and 4% bainitic microstructures but Figure 9 indicates a higher upper shelf COD for the structure containing 96% bainite. Figures 9 and 10 show that in the transition range the introduction of bainite into the martensitic structure raises the transition temperature. This shift in the transition from the curve of the martensitic structure is about 10°C for the structure containing 4% bainite, and about 20°C for the structure containing 96% bainite.

Metallographic and Fractographic Examination

42. Typical microstructures are shown in Figures 11 and 12. The "as-quenched" microstructure consisted of autotempered martensite, Figure 11. The 4% and 96% bainite microstructures are illustrated in Figures 12(a) and 12(b) respectively. The proportion of each constituent in the duplex microstructures appeared to be of the same order as that obtained by the systematic point counting technique using an optical microscope.

43. The fracture surfaces were examined in a Cambridge scanning electron microscope in order to establish the micro-mechanism of fracture. At + 23°C the fracture mode for all three microstructures was microvoid coalescence, Figure 16(a). In the transition region the predominant fracture mode was quasi-cleavage

with some area of microvoid coalescence, Figure 16(b). It was found that the quasi-cleavage facet size was much smaller in the microstructure containing 96% bainite, as is illustrated in Figure 17.

DISCUSSION

44. The simulated furnace heat treatment resulted in an austenite grain size of 50 μm, which is commensurate with that obtained midway between the grain coarsened and grain refined regions in real heat affected zones in HY130.

45. The results indicate a slight increase in the upper shelf COD values of the 96% bainite microstructure, Figures 9 and 10. This may be due to the lower tensile strength of the bainite, which could make it inherently tougher than the martensite when the fracture mode is microvoid coalescence.

46. The fracture mode for all three microstructures was microvoid coalescence in the temperature range between $-60^{\circ}C$ and $+25^{\circ}C$. In the transition region below about $-80^{\circ}C$, the micro mode of fracture was quasi-cleavage. It was observed that the addition of bainite to the martensitic structure raised the COD transition temperature, thereby indicating a reduction in the fracture toughness properties of the bainitic microstructure.

47. Thus it would appear that where fracture initiation occurs by microvoid coalescence, there is little difference in the relative fracture toughness of martensite and bainite. However where fracture initiation is by quasi-cleavage, there is a deterioration in the fracture toughness of the bainitic structure relative to the martensite. The formation of bainite could cause a lowering of toughness for several reasons. Firstly the carbide widths could be larger than in auto-tempered martensite, and secondly the width of bainite colonies could be larger on average than the average martensite colony width, thus leading to easier growth of cracks into adjacent colonies. A third factor could be the carbide shape, where carbide plates would be expected to fracture more readily than spheroidal carbides. Dolby and Knott (6) observed microcracks in bainitic colonies arresting in martensitic regions and concluded that it was likely that growth of nucleated cracks in bainitic carbides into ferrite is the controlling factor in initiation of fracture in these duplex microstructures, rather than the growth of cracks out of bainite colonies.

48. Comparison of Figures 13(a) and 15(a) indicate a coarser carbide precipitation in the bainitic structure than is evident in the autotempered lathes in the martensite. It is expected that the coarse acicular carbides in the bainite would initiate microcracks at lower stresses than the fine carbides in the autotempered martensite, resulting in a deterioration in the fracture toughness properties of the bainite.

CONCLUSIONS

49. An increase of 96% bainite in quenched martensitic HY130 tends to raise the upper shelf COD values slightly in the temperature range between $-60^{\circ}C$ and $+25^{\circ}C$. The microfracture mode in this temperature range for the three microstructures examined was found to be microvoid coalescence.

50. In the transition range, increasing amounts of upper bainite up to 96% raised the transition temperature by $20^{\circ}C$. The predominant fracture initiation mechanism in the transition region was found to be quasi-cleavage. The deterioration in the toughness of the bainite when fracture initiation is by a quasi-cleavage mechanism is probably due to the coarser carbide precipitation in the bainite.

51. These results indicate that where low temperature toughness is important, care should be taken to prevent the formation of bainite in the heat affected zone when welding HY130. Since the formation of bainite in the HAZ is possible when slow cooling rates occur after welding, lower preheat temperatures should be utilised to increase cooling rates.

ACKNOWLEDGEMENTS

52. Thanks are due to Mr I M Kilpatrick and Mr A R Morris, NCRE, for helpful discussion and criticism during the course of this work.

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TABLE 1

CHEMICAL COMPOSITION											
Element	C	Mn	P	S	Si	Ni	Cr	Mo	V	Cu	Al
HY130 Specification	0.12 max	0.60 to 0.90	.010 max	.015 max	0.20 to 0.35	4.75 to 5.25	0.40 to 0.70	0.30 to 0.65	0.05 to 0.10	- -	- -
Plate 5922 Analysis	0.13	0.79	.009	.009	0.25	4.87	0.52	0.51	0.06	0.05	.022

TIME AT 400°C (min)	PERCENTAGE BAINITE	VICKER'S PYRAMID NUMBER
Water Quench	0	399
0.5	4	387
1.0	53	374
2.0	60	363
3.0	74	356
4.0	86	341
5.0	96	338
PRIOR AUSTENITE GRAIN SIZE		50 μm

LAYOUT OF HEAT TREATMENT APPARATUS

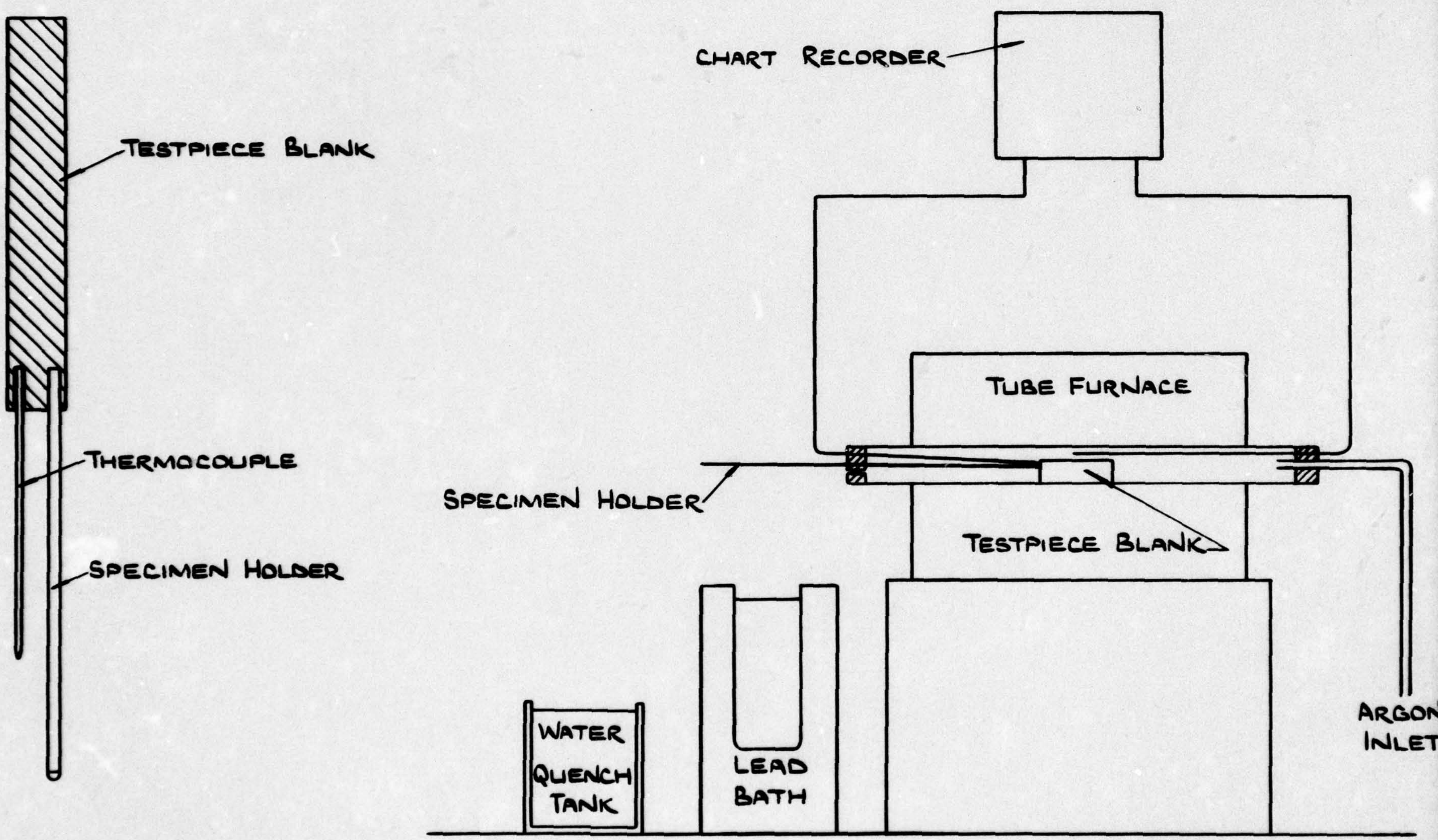


FIGURE 1

UNLIMITED

PERCENTAGE BAINITE FORMED AT 400° C

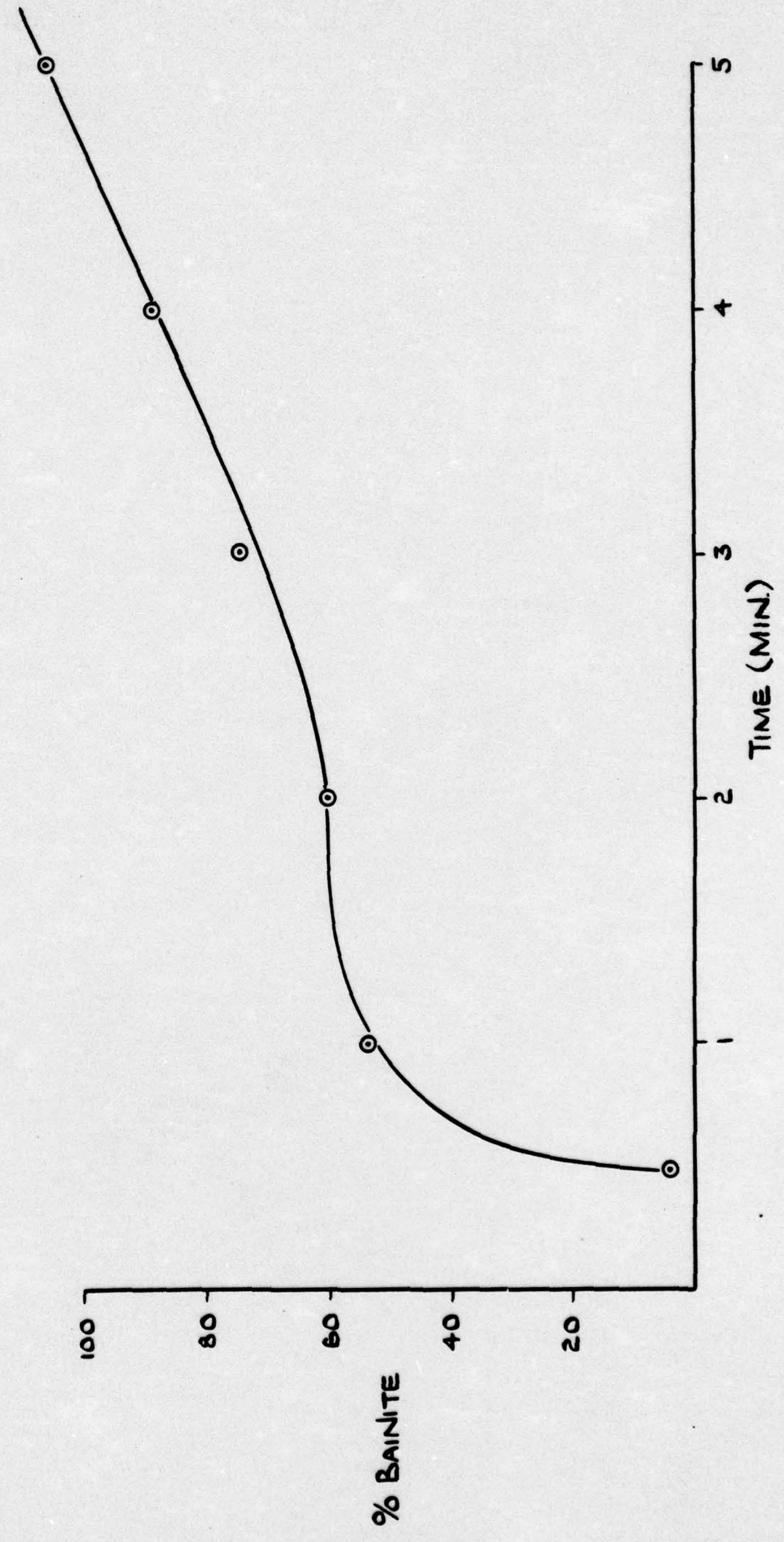


FIGURE 2

DIMENSIONS OF FRACTURE TOUGHNESS SPECIMENS

NOT TO SCALE
DIMENSIONS IN (mm)
NOTCH WIDTH 0·006"

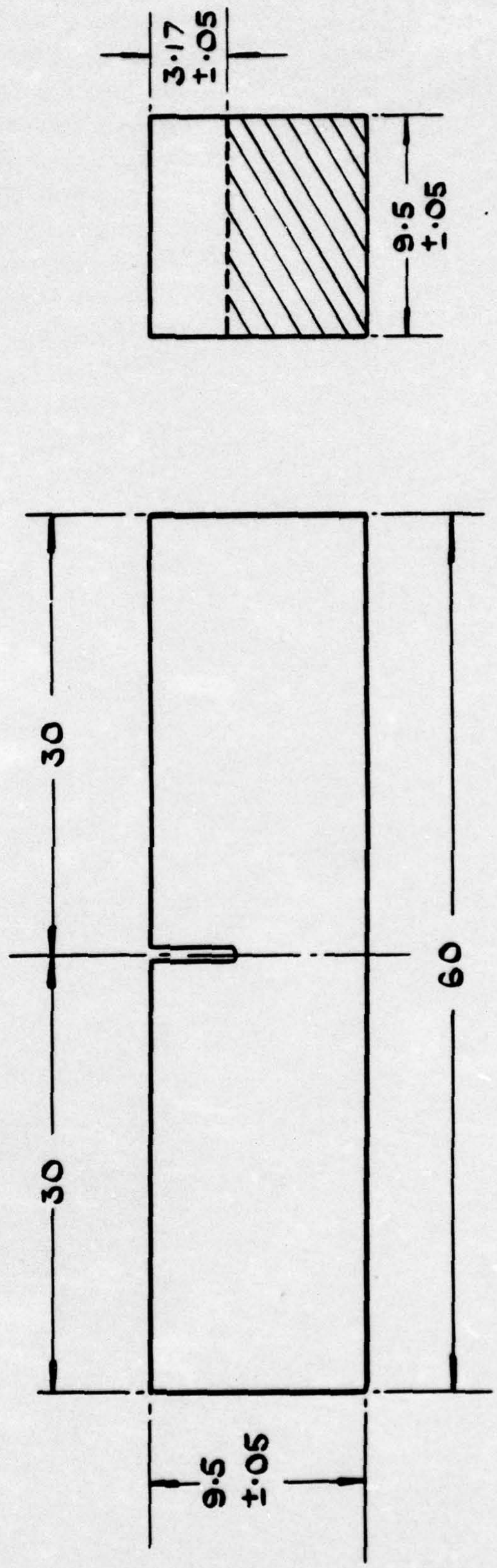


FIGURE 3

BEND TEST FIXTURE DESIGN

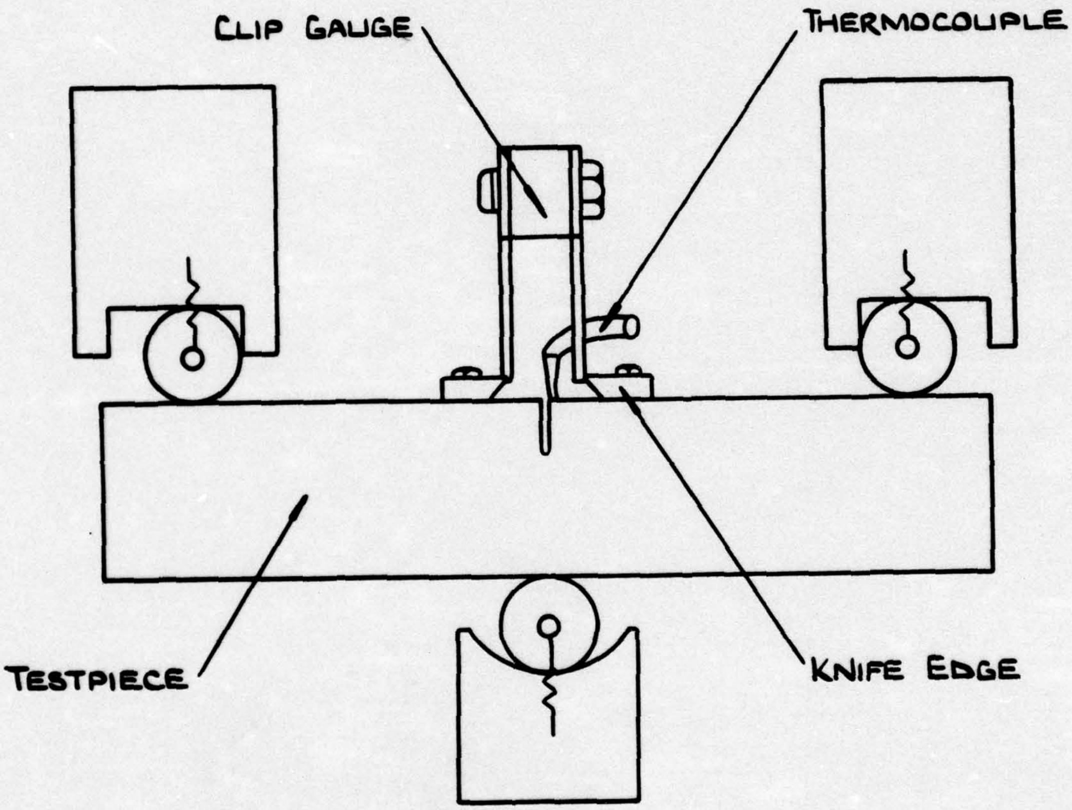


FIGURE 4

CHARACTERISTIC LOAD VERSUS C.O.D. RECORDS OBTAINED FOR A C.O.D TEMPERATURE TRANSITION CURVE.

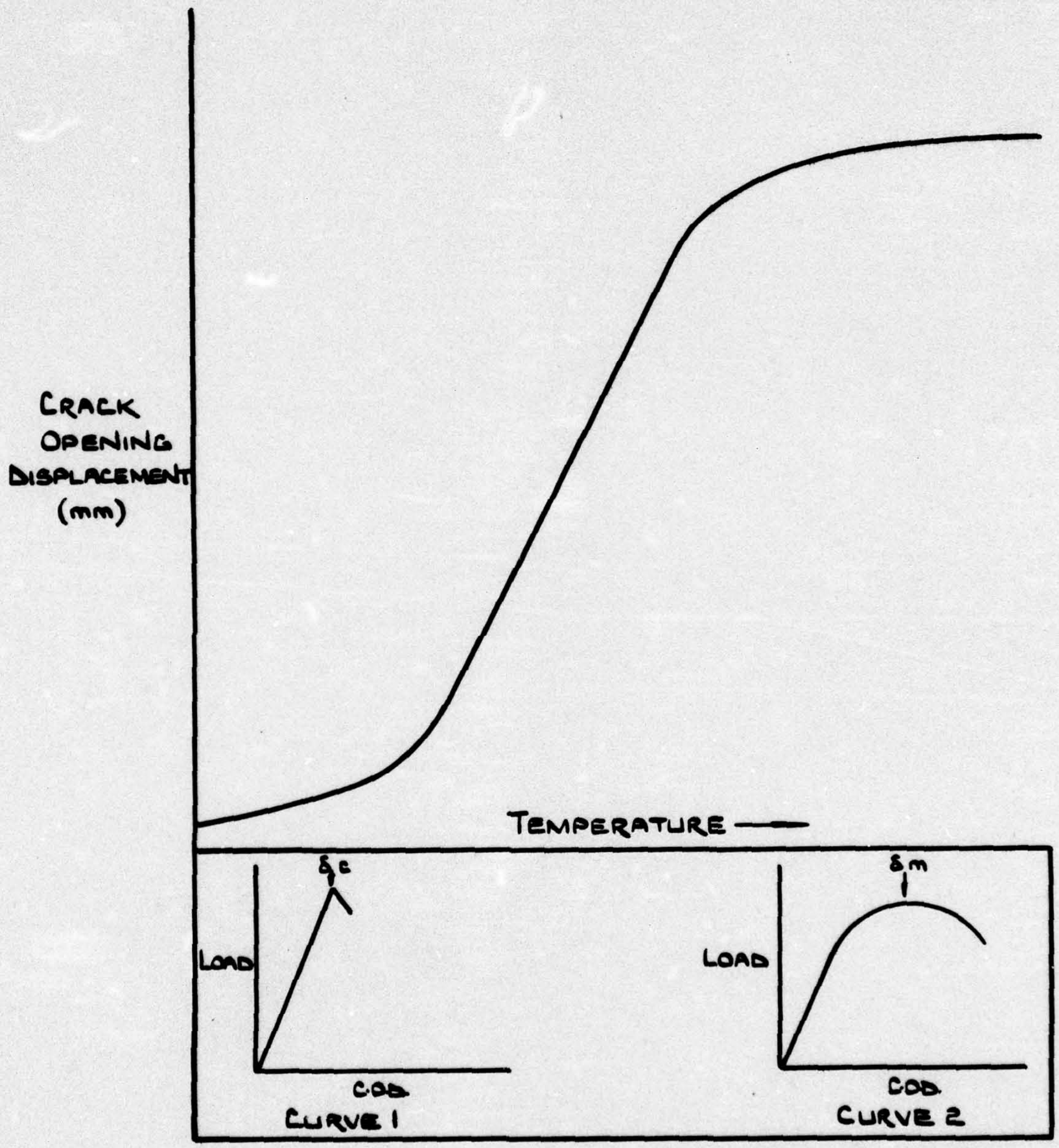


FIGURE 5

C.O.D. AS A FUNCTION OF TEMPERATURE FOR QUENCHED HY.130

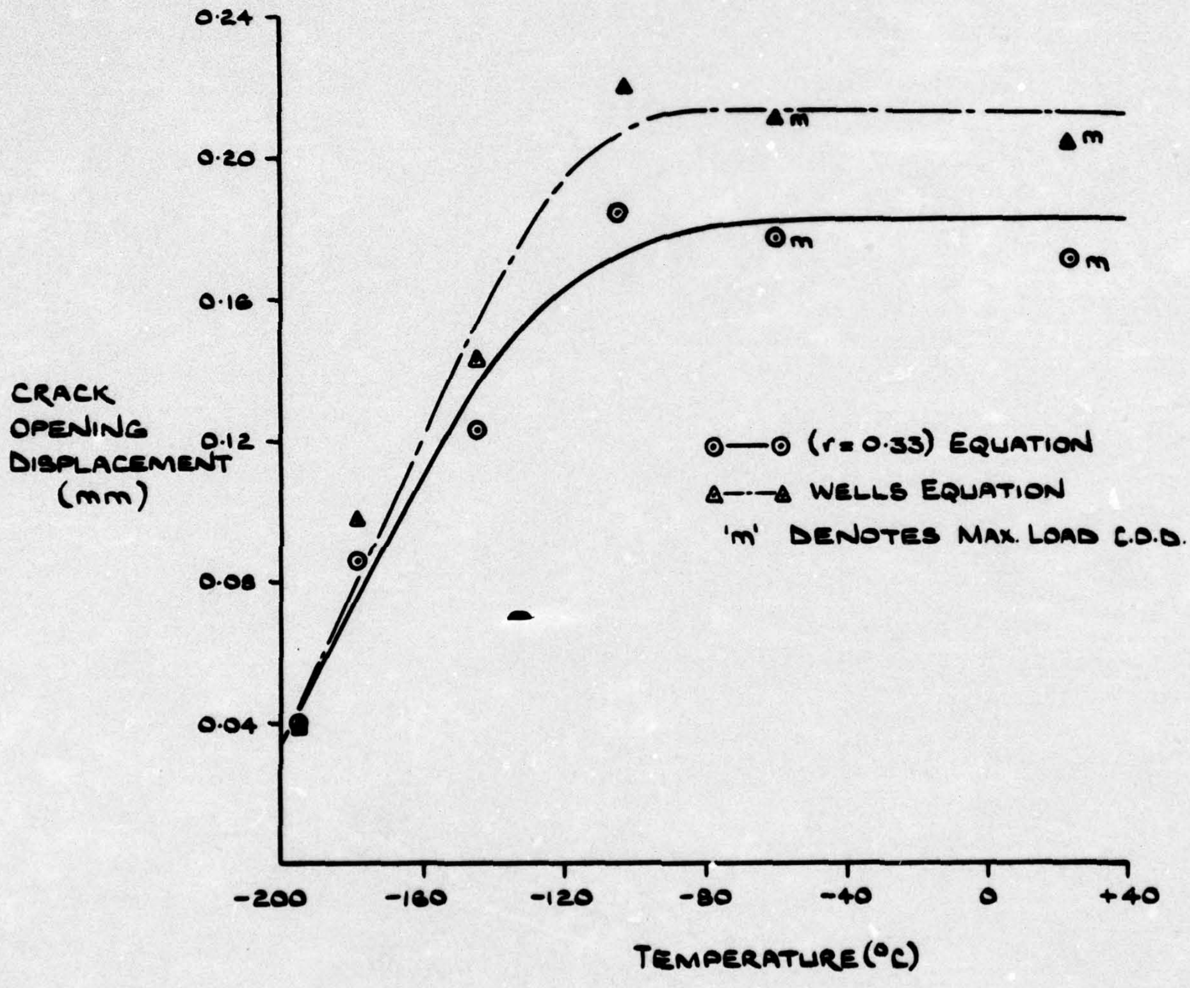


FIGURE 6

C.O.D AS A FUNCTION OF TEMPERATURE FOR HY.130 CONTAINING 4% BAINITE

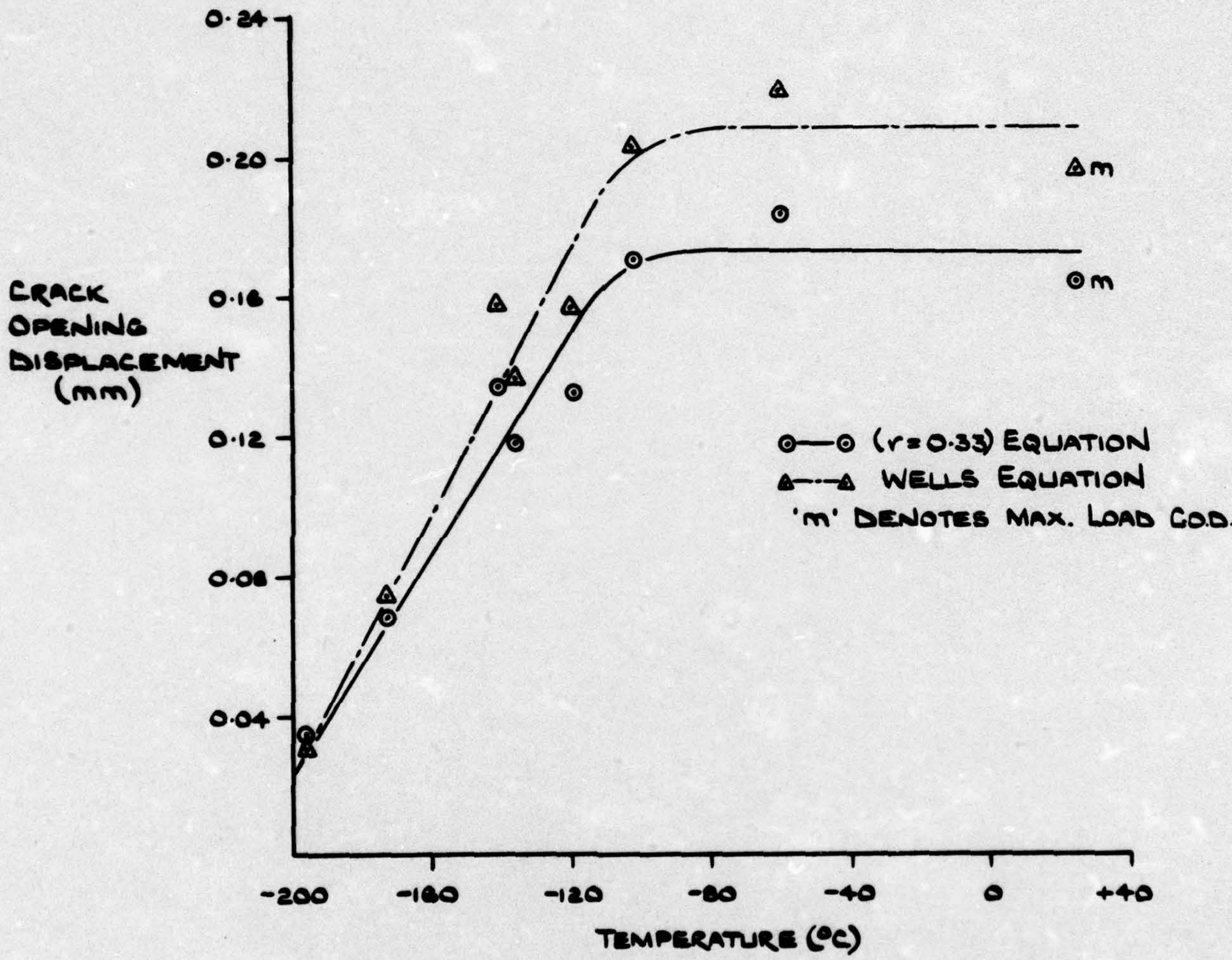


FIGURE 7

C.O.D. AS A FUNCTION OF TEMPERATURE FOR HY.130 CONTAINING 96% BAINITE

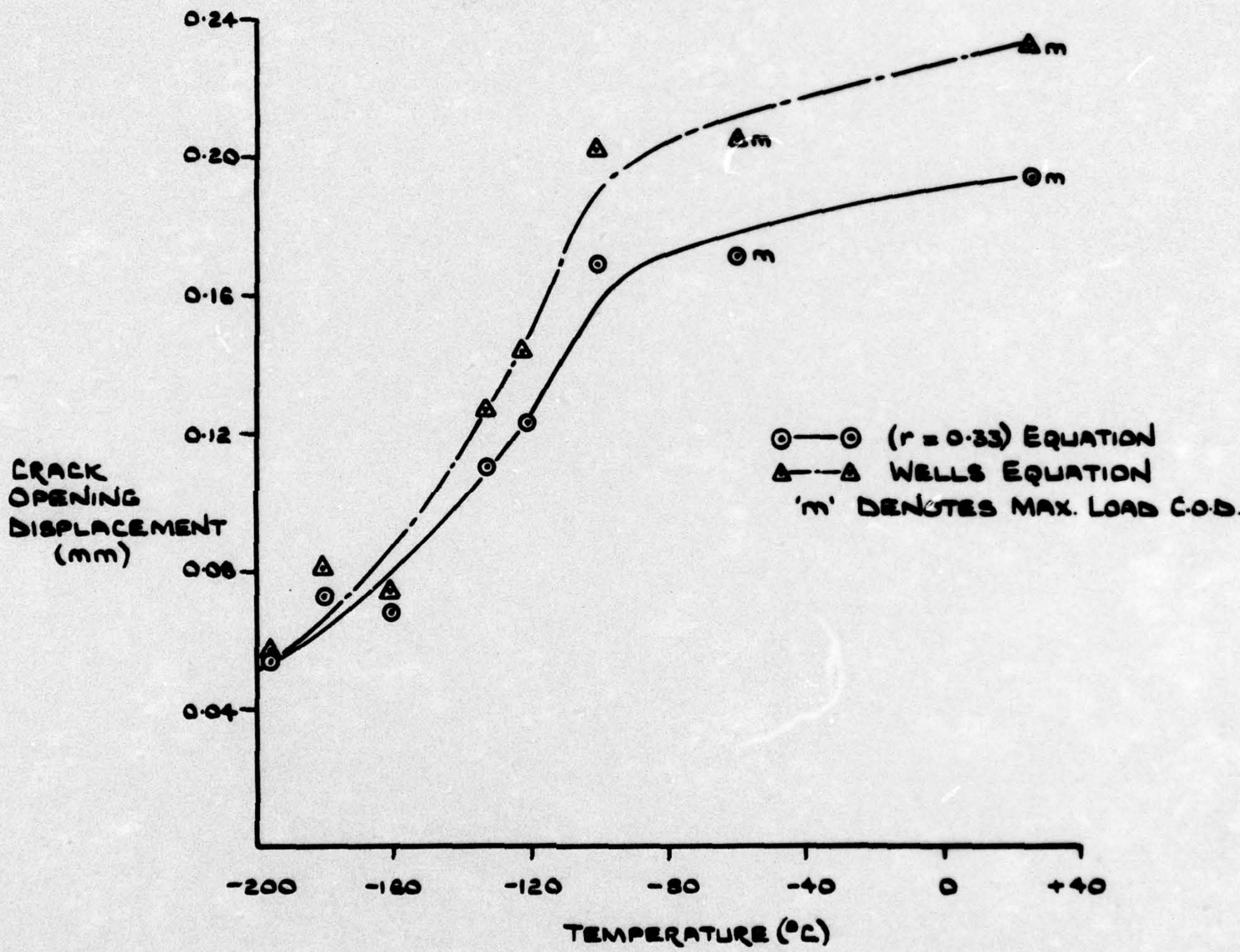


FIGURE 8

C.O.D (r = 0.33) AS A FUNCTION OF TEMPERATURE FOR HY 130 CONTAINING DIFFERENT AMOUNTS OF BAINITE

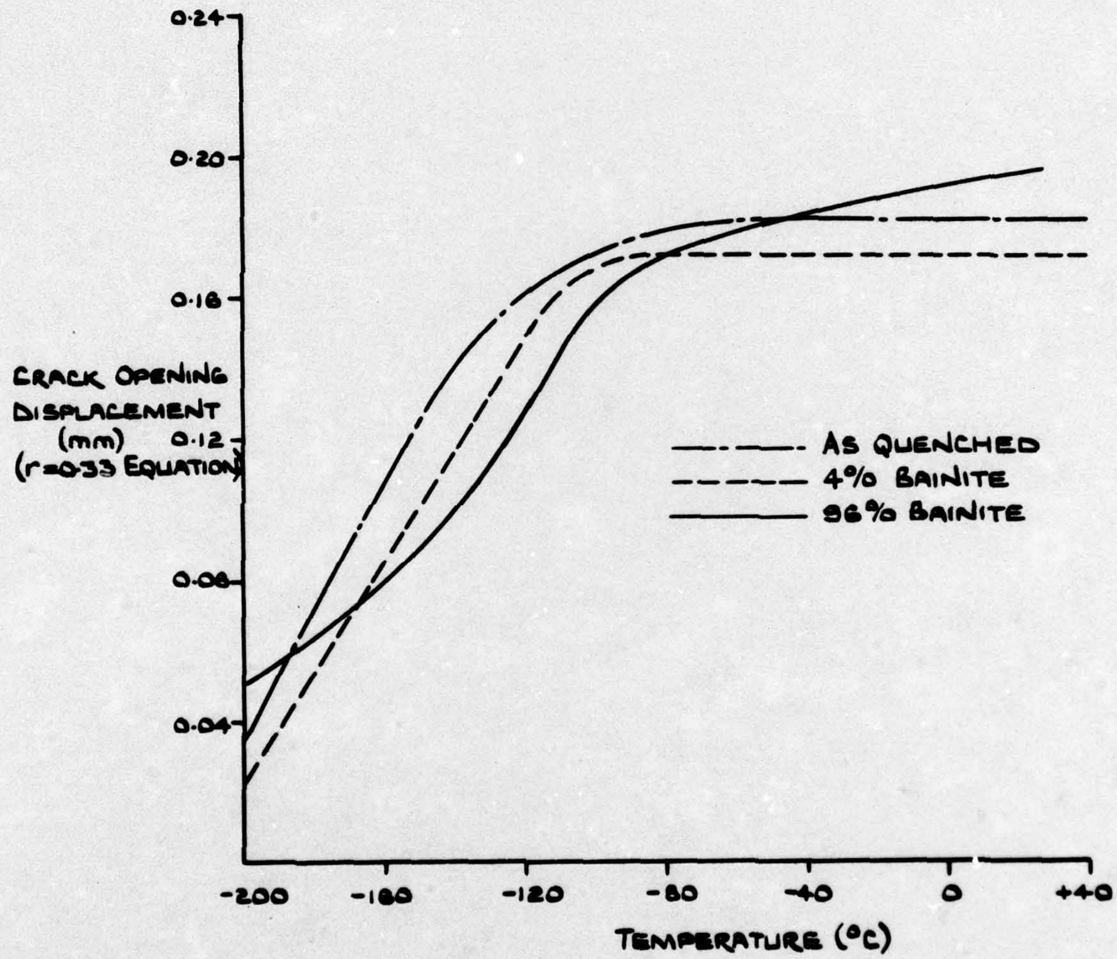


FIGURE 9

C.O.D (WELLS) AS A FUNCTION OF TEMPERATURE FOR H.Y. 130 CONTAINING DIFFERENT AMOUNTS OF BAINITE

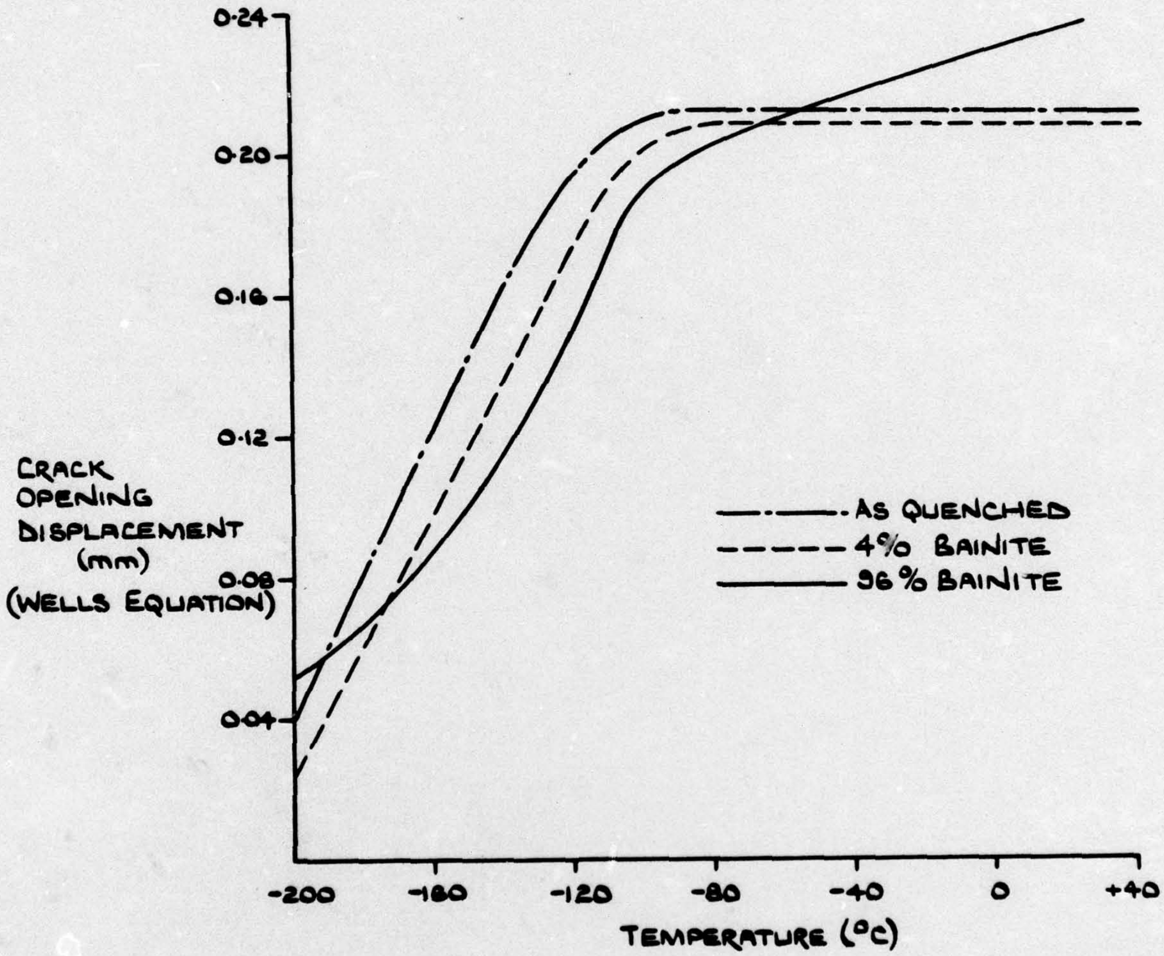
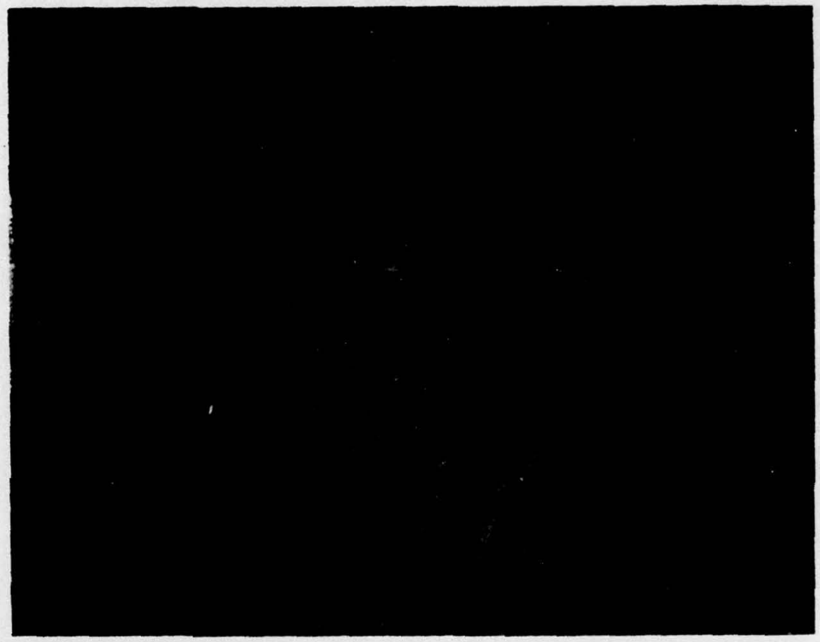


FIGURE 10

MARTENSITIC STRUCTURE FORMED ON QUENCHING INTO WATER FROM 1200°C



X325

FIGURE 11

MICROSTRUCTURES FORMED ON WATER QUENCHING FROM 1200°C & ISOTHERMALLY HEAT TREATING AT 400°C FOR (a) 30 SECS. & (b) 5 MINS.

(a)

x 325

(b)

x 325

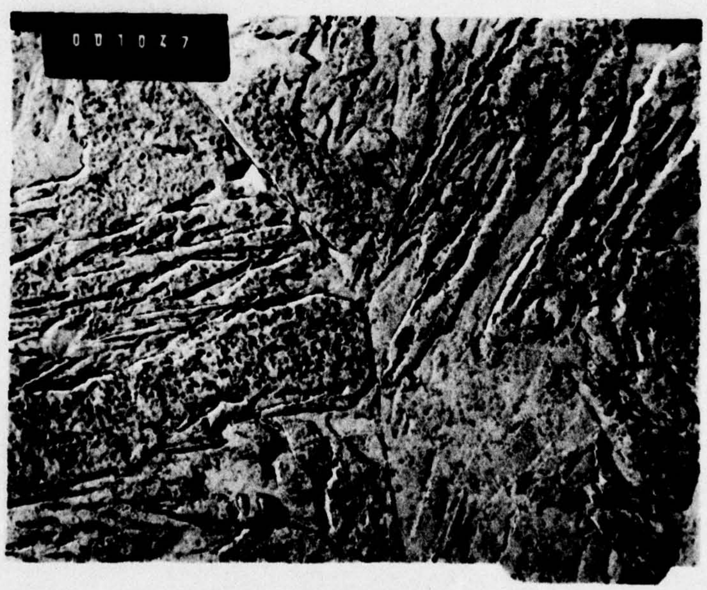
FIGURE 12

CARBON REPLICAS OF AS QUENCHED MICROSTRUCTURE (HY 130)

(a)

X 2256

(b)

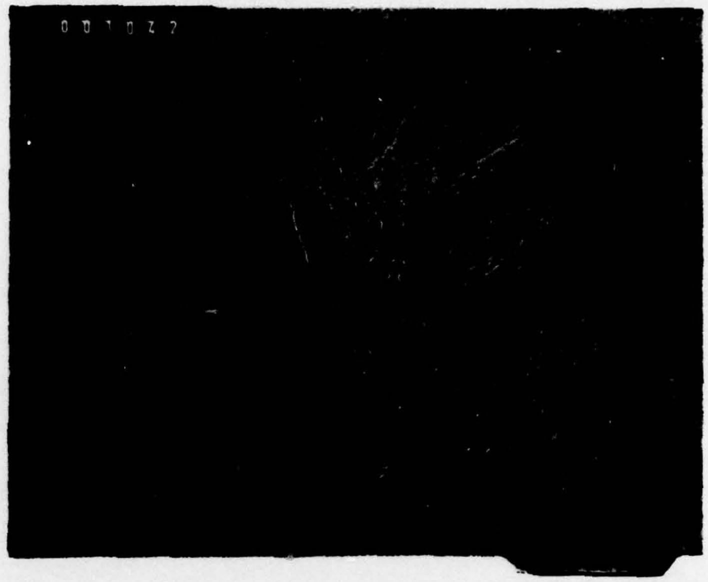


X 6768

FIGURE 13

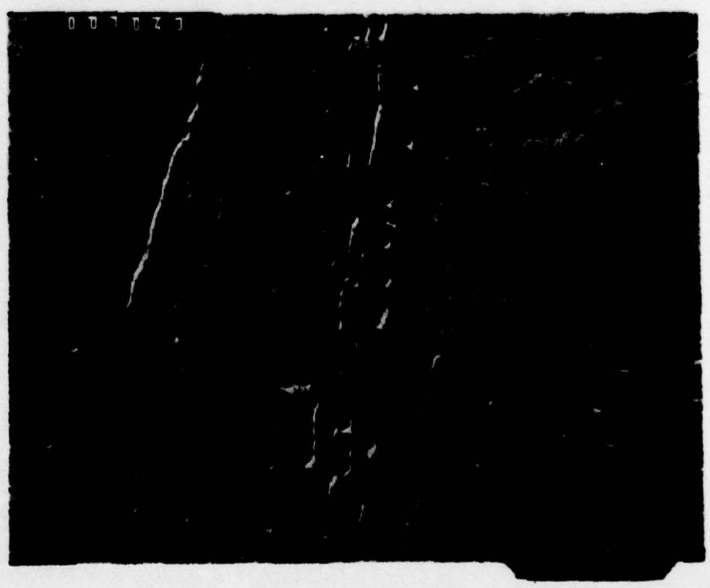
CARBON REPLICAS OF MICROSTRUCTURE CONTAINING 4% BAINITE (HY130)

(a)



X 2256

(b)

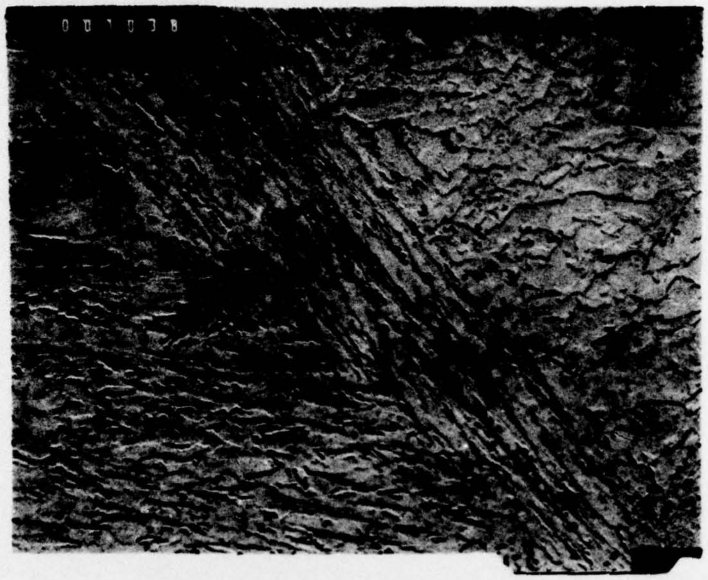


X 6768

FIGURE 14

CARBON REPLICAS OF MICROSTRUCTURE CONTAINING 96% BAINITE (HY 130)

(a)



X 2256

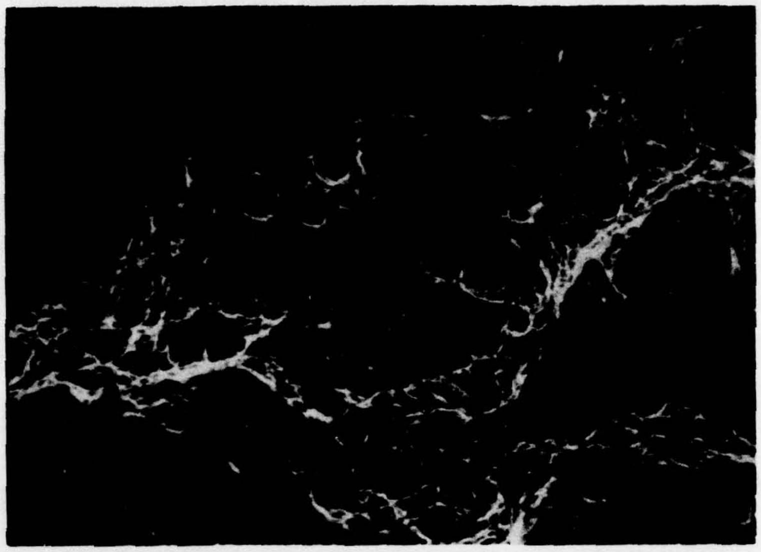
(b)

X 6768

FIGURE 15

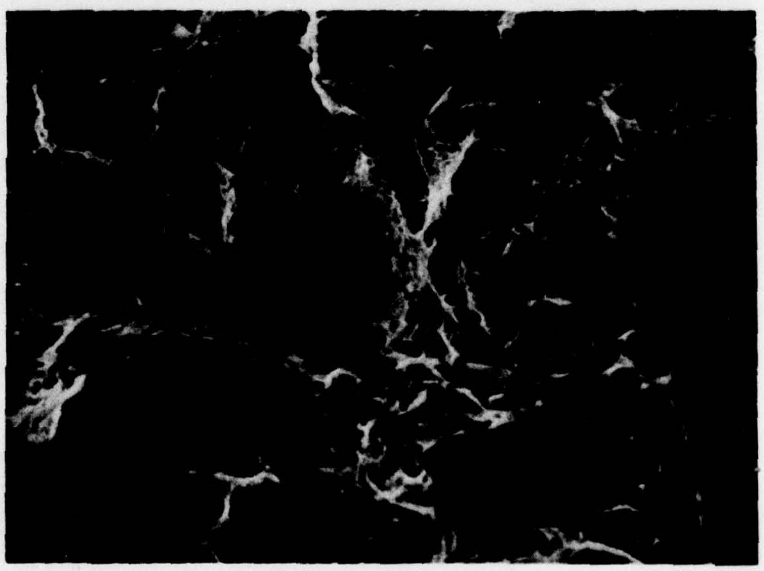
MICROSCOPIC FRACTURE MODES IN QUENCHED HY 130

TESTED AT +23°C
(a)



x 410

TESTED AT −179°C
(b)



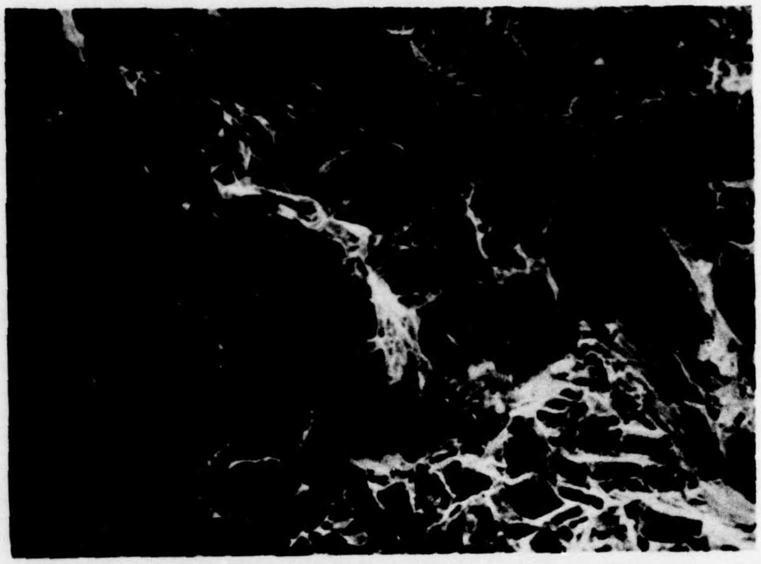
x 540

FIGURE 16

DIFFERENCE IN QUASI CLEAVAGE FRACTURES BETWEEN BAINITIC & MARTENSITIC MICROSTRUCTURES

4% BAINITE TESTED AT -120°C

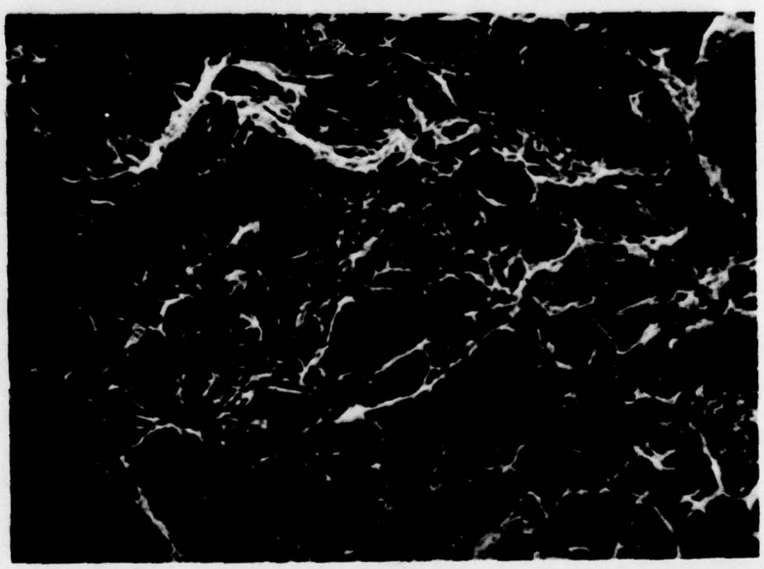
(a)



X 540

96 % BAINITE TESTED AT -122°C

(b)



X 540

FIGURE 17

DOCUMENT CONTROL SHEET

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5a. Sponsoring Agency's Code (if known)	**6a. Sponsoring Agency (Contract Authority) Name and Location**

7. Title

EFFECT OF PERCENTAGE BAINITE FORMED AT 400°C ON THE FRACTURE TOUGHNESS PROPERTIES OF AS-QUENCHED HY130

7a. Title in Foreign Language (in the case of translations)

7b. Presented at (for conference papers). Title, place and date of conference

8. Author 1. Surname, initials	**9a Author 2**	**9b Authors 3, 4...**	**10. Date**	**pp**	**ref**
Caudrey, A J				34	12

11. Contract Number	**12. Period**	**13. Project**	**14. Other References**

15. Distribution statement

[redacted]

Descriptors (or keywords)

Fracture toughness
Bainite
HY130

Abstract

The comparative fracture toughness properties of bainite and martensite in a low alloy high yield strength structural steel have been investigated using the crack opening displacement (COD) technique of general yielding fracture mechanics.

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